Hydrogen-Induced Cracking along the Fusion Boundary of Dissimilar Metal Welds

The susceptibility of dissimilar austenitic/ferritic combinations to hydrogen-induced cracking near the fusion boundary has been investigated

BY M. D. ROWE, T. W. NELSON AND J. C. LIPPOLD

Introduction

Dissimilar metal welds are used extensively in the power generation, petrochemical and heavy fabrication industries. Numerous instances of cracking along the dissimilar metal fusion boundary have been reported, particularly in cladding applications where a corrosion-resistant austenitic alloy is applied to a ferritic structural steel. Often this cracking, or disbonding, has been associated with exposure to hydrogen in service and, as a result, the mechanism has been described by various authors as a form of hydrogen-induced cracking (Refs. 1–13). This type of cracking has been reproduced in the laboratory by exposing austenitic cladding to hydrogen, either in an autoclave or by cathodic charging (Refs. 1–3, 7, 8, 11–13).

In practice, however, this form of cracking has occurred during fabrication, prior to exposure to a hydrogen environment. The fact that disbonding can occur without prolonged exposure to hydrogen in service suggests that either hydrogen is not necessary for disbonding to occur, or hydrogen absorbed during welding can cause cracking near the dissimilar metal fusion boundary.

The fusion boundary microstructure in dissimilar welds often possesses some unique features. Normal epitaxial nucleation during solidification along the fusion boundary gives rise to grain boundaries that are continuous from the base metal into weld metal across the fusion boundary. These boundaries are roughly perpendicular to the fusion boundary and have been referred to as “Type I” boundaries. In dissimilar welds, where an austenitic weld metal and ferritic base metal exist, a second type of boundary that runs roughly parallel to the fusion boundary is often observed. These boundaries are roughly parallel to the fusion boundary and have been referred to as “Type II” boundaries. In dissimilar welds, where an austenitic weld metal and ferritic base metal exist, a second type of boundary that runs roughly parallel to the fusion boundary is often observed. This has been referred to as a “Type II” boundary (Ref. 6). These boundaries typically have no continuity across the fusion boundary to grain boundaries in the base metal. Several investigators have reported that hydrogen-induced disbonding typically follows Type II grain boundaries (Refs. 1-4, 7, 8, 12, 13). The disbonding phenomenon that occurs following fabrication and prior to service has also been associated with these Type II boundaries.

An additional complication in austenitic/ferritic dissimilar welds is the dramatic transition in composition and
microstructure that occurs adjacent to the fusion boundary. This transition can be illustrated using the Schaeffler Constitution Diagram (Ref. 14). If a tie line is drawn on this diagram (Fig. 1) from a ferritic steel base metal to an austenitic stainless steel filler metal (such as Type 308 or 309LSi) or a nickel-based filler metal (such as ERNiCr-3), it can be seen that intermediate compositions along the tie line between the end points will promote martensitic and austenitic plus martensitic microstructures. In practice this transition occurs over a very short distance (less than 1 mm) from the fusion boundary into the weld metal, and results in a localized martensitic band along the fusion boundary. Cracking has been reported in the martensitic transition zone near the fusion boundary (Refs. 3, 12). Often, the Type II boundaries described previously reside in this martensitic region.

In order to more carefully study the effect of fusion boundary microstructure on hydrogen-induced cracking in dissimilar welds, a number of dissimilar welds were made using various austenitic filler metals using both pure argon and Ar-6%H₂ shielding gases. The objective of these trials was to produce hydrogen-induced cracking in dissimilar metal welds by the addition of hydrogen during welding, and to characterize the microstructures in which cracking occurs. An understanding of the susceptibility of dissimilar metal fusion boundary microstructures to hydrogen-induced cracking will contribute to an understanding of defect formation in dissimilar metal welds and aid in development of sound welding procedures.

### Experimental Procedure

#### Materials

The filler materials selected for this investigation are commonly used in industry for dissimilar metal welding. Type 308, 309LSi and ERNiCr-3 filler metals were selected to cover a range of compositions and microstructures. A36 steel was selected as the base metal. The chemical compositions of the materials are listed in Table 1.

<table>
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<tr>
<th>Element</th>
<th>A36</th>
<th>ER308</th>
<th>Er309LSi</th>
<th>ERNiCr-3</th>
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<tr>
<td>Fe</td>
<td>bal</td>
<td>bal</td>
<td>bal</td>
<td>1.33</td>
</tr>
<tr>
<td>Cr</td>
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<td>20.51</td>
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<tr>
<td>Mo</td>
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<tr>
<td>Cu</td>
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<td>0.24</td>
<td>0.20</td>
<td>0.09</td>
</tr>
<tr>
<td>Ti</td>
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<td>—</td>
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<td>0.020</td>
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<td>N</td>
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<td>0.110</td>
<td>—</td>
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<td>0.019</td>
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<td>—</td>
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<tr>
<td>S</td>
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<td>0.022</td>
<td>0.014</td>
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<td>Ni_{eq}</td>
<td>5.59</td>
<td>11.97</td>
<td>15.25</td>
<td>76.18</td>
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(a) Schaeffler equivalents (Ref. 3).

#### Welding Procedures

The gas tungsten arc welding (GTAW) process was selected because it allows for close control of dilution and the addition of hydrogen through the shielding gas. Shielding gases consisting of pure argon and Ar-6%H₂ were used. Both a multipass and single-pass welding procedure were developed to assess the effect of hydrogen introduction through the shielding gas.

The multipass weld procedure joined two 0.75 x 4 x 12 in. (19 x 102 x 305 mm) plates of A36 steel. Welding was performed along the 12-in. dimension. The plates were restrained in a heavy fixture to simulate actual high-restraint fabrication conditions. Both ER308 and ERNiCr-3 filler metals were used to fill a standard V-groove joint geometry. The included angle of the groove was 60 deg, and 16–20 passes were required to fill the joint with the welding conditions listed in Table 2. Following welding, the weldment was left rigidly restrained for up to four days, then inspected for cracking using side-bend tests, and by sectioning and metallography.

A single-pass procedure followed by application of augmented strain by bending was used to allow for greater control of applied strain and to minimize the elevated temperature diffusion of hydrogen that occurs during a multipass procedure. The single-pass welds were deposited in a V-groove using the conditions listed in Table 2. Three percent strain was applied by bending over a fixed-radius die block in either the longitudinal or transverse direction with respect to the weld. Each test plate contained two welds deposited side by side.
one made with pure argon, the other with Ar-6%H₂ shielding gas. First, the pure argon weld was deposited, then the plate was cooled in water. Secondly, the Ar-6%H₂ weld was deposited and the plate was cooled in water again. Finally, the welds were ground flush with the surface of the plate and the augmented strain was applied within 5 min of extinguishing the arc. Crack propagation across the surface of the weld was monitored visually on samples subjected to longitudinal strain.

Addition of hydrogen to the shielding gas caused an increase in dilution (deeper penetration) for a given current level. It was therefore necessary to use a lower current with the Ar-6%H₂ shielding gas to achieve a similar dilution as welds made with pure argon. Dilution was controlled by making minor adjustments to the current levels listed in Table 2. Metallographic cross sections were prepared to assure that the dilution levels were similar.

**Microstructural Characterization**

Microstructural characterization was performed using optical metallography, microhardness, energy dispersive X-ray spectroscopy (EDS) and dilution measurements. Because of the range of compositions and microstructures a number of chemical etchants were used, including 10% chromic acid (electrolytic), 5 g FeCl₃/90 mL methanol/2 mL HC1 (electrolytic) and 4% nital.

A ferrofluid technique was used to provide contrast between martensite and austenite in the weld metal. Ferrofluid is a colloidal suspension of ferromagnetic iron oxide particles. The ferrofluid is applied with an eye dropper onto the surface of the sample in either the as-polished condition or after previous chemical etching, and then gently rinsed in a bath of petroleum ether to remove the excess ferrofluid. The sample was then removed from the petroleum ether and allowed to dry. The remnant magnetism of any ferromagnetic phases (ferrite and martensite) attracts the iron oxide particles, but leaves the paramagnetic phase (austenite) free of particles. The difference in ferrofluid deposition produces a color contrast between ferromagnetic phases and paramagnetic phases when viewed in an optical microscope. Ferrofluid has been successfully used to provide contrast between ferrite and austenite in duplex stainless steels (Ref. 15) and between austenite and strain-induced martensite in wrought austenitic stainless steels (Ref. 16).

Microhardness was used to study the transition in microstructure at the fusion boundary. A diamond pyramid indenter was used in conjunction with both 10- and 100-g loads. Plots of hardness vs. distance from the fusion boundary were used to support metallographic observations and determine the width of the transition region.

Weld metal dilution measurements were made on the single-pass welds in order to relate the predicted composition to the microstructure of the weld metal from the Schaeffler diagram. To determine dilution, an image of the weld cross section was scanned into a computer, and a graphics software package was used to measure the weld nugget area relative to the original V-groove dimension. The filler metal dilution is then given by the area of base metal melted divided by the nugget area.

### Table 2 — Welding Conditions

<table>
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<tr>
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<th>Multipass Welds</th>
<th>Single-Pass Welds</th>
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<tr>
<td>Root Passes</td>
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<td>250</td>
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<tr>
<td>Fill Passes</td>
<td>275</td>
<td>225</td>
</tr>
<tr>
<td>Argon</td>
<td>12.5</td>
<td>12.5</td>
</tr>
<tr>
<td>Argon+6%H₂</td>
<td>13.5</td>
<td>13.5</td>
</tr>
<tr>
<td>Current (A)</td>
<td>250</td>
<td>250</td>
</tr>
<tr>
<td>Voltage (V)</td>
<td>106(200)</td>
<td>225(99)</td>
</tr>
<tr>
<td>Wire Feed Speed (in./min)</td>
<td>85(99)</td>
<td>225(99)</td>
</tr>
<tr>
<td>Travel Speed, (mm/s)</td>
<td>7(3)</td>
<td>7(3)</td>
</tr>
<tr>
<td>(mm/s)</td>
<td>7(3)</td>
<td>7(3)</td>
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</tbody>
</table>

(a) Wire feed speed for 0.045 in. (1.14 mm) diameter wire.

Fig. 3 — Plan view, ER308 filler metal, transverse strain, Ar-6%H₂ shielding gas, chromic acid/nital etch, showing cracking running parallel to the fusion boundary.

Fig. 4 — Plan view, same weld that appears in Fig. 2 treated with ferrofluid on a polished surface. A — Cracking occurred in a region colored strongly by ferrofluid, indicating the presence of martensite; B — A+F indicates an austenite plus skeletal ferrite microstructure, A+M indicates an austenite plus martensite microstructure.
Results

Multipass Weld Procedure

Side-bend tests and metallography revealed no cracking in any of the multipass welds made with ER308 and ErNiCr-3 filler metals. Metallography revealed a martensitic transition region near the fusion boundary in both welds with hardness in excess of 400 HV. More than an hour was required to complete a multipass weld, and upon completion, the plate temperature was in excess of 500°F (260°C). The time spent at elevated temperature was probably sufficient to allow hydrogen to diffuse away from the weld area, leaving an insufficient concentration to cause cracking. It is also possible that the residual stress from welding was insufficient to promote cracking. This procedure was abandoned in favor of a single-pass technique with application of augmented strain.

Single-Pass Weld Procedure

Cracking was observed in welds made with each of the three filler metals using Ar-6%H₂ shielding gas. Cracking was not observed in any of the welds made with pure argon shielding gas. In welds made with Ar-6%H₂ shielding gas, cracking occurred within 5 to 30 min after augmented strain was applied. In welds made with ER308 filler metal with dilution near 40%, cracks propagated throughout the weld metal. When the dilution was reduced to 30%, cracking was confined to a narrow region of the weld metal within 1 to 1.5 mm of the fusion boundary.

With ER309LSi filler metal at 44% dilution, cracking was also confined to within 1 to 1.5 mm from the fusion boundary. With ERNiCr-3 filler metal at 38% dilution, cracking was not visible to the naked eye, but microscopic cracks in a narrow weld metal band 100 to 150 µm of the fusion boundary were revealed by metallography.

ER308 Weld Deposits

The ER308 filler metal experienced more severe cracking for a given level of dilution than the other two filler metals. Figure 2 shows the appearance of cracking in a weld made with the ER308 filler metal and argon-6%H₂ shielding gas after longitudinal bending. This weld had a relatively low dilution of 30%, promoting the formation of austenite plus ferrite in the bulk weld metal and partially martensitic microstructures in regions near the fusion boundary. Note that the cracks in

Fig. 5 — Plan view, Type 309LSi filler metal, longitudinal strain, 44% dilution, Ar-6%H₂ shielding gas, chromic acid/nital etch. A — Cracking and hardness traverse near the fusion boundary; B — detail of crack tip that appears in (A) showing propagation path.

Fig. 6 — Plan view, same weld that appears in Fig. 5 treated with ferrofluid after chromic acid/nital etch. Martensitic regions are colored brown, austenitic regions remain white. A — Alternating bands of austenite and martensite are visible in the weld metal; B — crack near the fusion boundary. High hardness in the darker regions reflects the presence of martensite.
Fig. 2 are associated with light-etching bands in the weld metal. Figure 3 shows cracks propagating parallel to the fusion boundary in a sample subjected to transverse strain. Figure 4 shows the microstructure of the weld, which appears in Fig. 2 as revealed by ferrofluid applied to the polished surface. The region near the fusion boundary where the cracking occurred is strongly colored by ferrofluid (blue or purple) in Fig. 4A and has an average hardness of 476 HV, which is significantly higher than the bulk weld metal average hardness of 311 HV.

The region near the center of Fig. 4B labeled A+F has a microstructure of austenite plus skeletal ferrite, which is typical of ER308 weld deposits. The austenite is white or light brown while the ferrite is colored. The region labeled A+M has a microstructure of austenite plus martensite. Hardness indentations in Fig. 4B indicate that the region colored by ferrofluid has a significantly higher hardness than the white region, which is consistent with the presence of martensite. When compared to Fig. 2, it was found that the light etching bands where cracking initiated corresponded to the A+M regions in Fig. 4. It can be concluded from the high hardness and coloration by ferrofluid that the cracking occurs in regions containing martensite.

**ER309LSi Weld Deposits**

Type 309LSi filler metal exhibited less severe cracking than Type 308 for a given dilution. Figure 5 shows cracking in a region of high hardness near the fusion boundary in a weld made with Ar-6%H₂ shielding gas. The crack extends 600 µm away from the fusion boundary. Figure 5B shows the propagation path at the tip of the crack in Fig. 5A. Both intergranular and transgranular propagation are evident. This is consistent with Savage, et al. (Ref. 17), who observed both intergranular and transgranular propagation in experiments involving direct observation of hydrogen-induced cracking. The chromic acid etch does not clearly reveal the austenite plus martensite structure through which the crack propagates. The average hardness in the crack region, as determined from the traverse shown in Fig. 5A, is 431 HV, while the bulk weld metal hardness is 239 HV.

Figure 6 shows the same weld that appears in Fig. 5 treated with ferrofluid after being etched with chromic acid and nital. Martensitic regions are brown while austenitic regions remain white. Hardness indentations indicate a significantly higher hardness in the regions colored by ferrofluid, which is consistent with the presence of martensite. Cracking is visible in the band of high hardness martensite in the weld metal adjacent to...
ERNiCr-3 weld metal experienced the least severe cracking of the three filler metals, and formed the least amount of martensite at comparable dilutions. Figure 8 shows a microscopic crack approximately 110 μm long in a region of martensite, which appears a darker gray color than the bulk weld metal. The average hardness in the gray (martensitic) region of Fig. 8 is 462 HV, while the average hardness in the adjacent austenite is 206 HV. When treated with ferrofluid, the gray region was colored while the austenitic weld metal remained white. When treated with a ferric chloride etch (Fig. 9), a lathy microstructure was revealed in the hard gray region near the fusion boundary, which is consistent with the presence of martensite. The ferric chloride etch was also effective at revealing Type II grain boundaries, as indicated by the arrows in Fig. 9, which were not revealed by the chromic acid etch.

A composition profile was measured across the martensitic gray region, which appears in Fig. 8, using SEM/EDS. Local percent filler metal dilution was calculated using the results of the EDS measurements for Ni and Cr, as shown in Fig. 10. The Schaeffler diagram predicts that the minimum dilution from an 36 base metal to form martensite in ERNiCr-3 weld metal is 78% and the minimum dilution to form a fully martensitic microstructure is 84%, as shown in Fig. 1. It can be seen from Fig. 10 that the Schaeffler diagram predicts an almost fully martensitic microstructure in the gray region where the crack appears in Fig. 8.

**Discussion**

Cracking was observed in single-pass welds made with Ar-6%H₂ shielding gas using all three filler metals. Cracking was not observed in welds made with pure argon; therefore it can be concluded that the cracking was hydrogen induced.

Cracking was always associated with regions of martensite near the fusion boundary. The ferrofluid color metallurgy technique revealed the cracks occurred in regions containing a ferromagnetic constituent, which is consistent with the presence of martensite. Microhardness indentations revealed that the regions where cracking occurred had a hardness significantly higher than that of the austenitic weld metal, which is also consistent with the presence of martensite. An electrolytic etchant consisting of 5 g FeCl₃, 90 mL methanol and 2 mL HCl revealed a martensitic microstructure near the fusion boundary in the weld made with ERNiCr-3 filler metal, as shown in Fig. 9.

Welds made with ERNiCr-3 and ER309LSI filler metals exhibited less cracking than welds made with 308 filler metal. According to the Schaeffler diagram (Fig. 1), the minimum base metal dilution necessary to form martensite in these filler metals is 16% for ER308, 33% for ER309LSI and 78% for ERNiCr-3. For a given dilution, welds made with ER308 filler metal contained the most martensite and experienced the most severe cracking. Low dilution welds made with ER309LSI and ERNiCr-3 can only form crack-susceptible martensite in select locations near the fusion boundary where the local dilution is higher than that of the bulk weld metal.

Locally increased dilution near the fusion boundary is most likely the result of a stagnant fluid layer in the weld pool contacting the solid base metal. The martensitic regions may form by a mechanism similar to the unmixed zone described by Savage and Szekeres (Ref. 18). The martensitic regions differ from an unmixed zone in that they do not have the same composition and microstructure as the base metal. A true unmixed zone seems unlikely to form in a dissimilar metal weld because of the strong composition gradient between the bulk weld metal and the stagnant fluid layer while both are in the liquid state. Duvall and Owczarski concluded that diffusion from the bulk weld metal into the unmixed zone will both be liquid is feasible considering the conditions of time, temperature and distance in an arc weld (Ref. 19).

Although a lower heat input was used for the Ar-6%H₂ welds than those made with pure argon, the welds with hydrogen addition had a somewhat higher dilution. Dilution levels for some of the pairs of welds (pure argon/Ar-6%H₂) presented in the results section are 28%/30% for ER308, 33%/44% for ER309LSI and 32%/38% for ERNiCr-3. The authors believe that cracking was caused by hydrogen addition and not by increased dilution. The susceptible microstructure, which is the band of high hardness martensite near the fusion boundary, was present in all of the welds regardless of whether they were made with pure argon or Ar-6%H₂. The fact that cracking only occurred in welds made with hydrogen addition, even though the susceptible microstructure was present in both types of welds, indicates that the cracking was hydrogen induced.

Cracking along Type II grain boundaries is frequently reported in the literature; however, it was not observed in this investigation. Most of the experiments in this investigation involved longitudinal bending, which does not place tension across the Type II grain boundaries. Longitudinal bending is effective at determining the width of the crack-susceptible region near the fusion boundary and the range of hardness in which cracks will propagate, but not effective at determining crack susceptibility of the Type II boundaries.

**Practical Implications**

The results of this study have shown that hydrogen introduced during welding can lead to hydrogen-induced cracking.
in dissimilar welds between austenitic filler metals and ferritic base metals. Hydrogen in the welding arc is detrimental in two ways: 1) it increases dilution by the carbon steel base metal, increasing the amount of martensite formed, and 2) it interacts with martensite under stress to cause cracking. The incidence of cracking was most pronounced in single-pass welds. This suggests that the use of multipass techniques or thermal treatments that allow for hydrogen diffusion will minimize cracking susceptibility. The use of low-hydrogen practice with dissimilar metal welds is also suggested.

Cracking was always associated with hard, martensitic regions adjacent to the fusion boundary; therefore, minimizing the compositional regime within which martensite can form will reduce susceptibility. As this study has shown, filler metal selection can have a profound influence on the tendency to form martensite, with ER308 showing the strongest tendency and ERNiCr-3 the weakest. Reducing base metal dilution will also reduce the tendency to form martensite in both the bulk weld metal and along the fusion boundary, as illustrated by the Schaeffler diagram.

Finally, it is advisable to avoid stress concentrations at the fusion boundary, such as undercut or a sharp toe angle, because they locally increase applied stress and help to initiate hydrogen-induced cracking. Since the most susceptible microstructure forms in the vicinity of the fusion boundary, elimination of stress concentrators at the toe or root of the weld is imperative.

Conclusions

1) Hydrogen-induced cracking was observed in high-hardness martensite (400 to 550 HV) near the dissimilar metal fusion boundary between ER308, ER309LSi and ERNiCr-3 filler metals and A36 steel base metal.

2) Cracking was observed in welds made with Ar-6%H₂ shielding gas, but not in welds made with pure argon shielding gas.

3) A ferrofluid color metallography technique revealed that cracking was always confined to regions containing martensite.

4) The severity of cracking observed in the filler metals tested was a function of the minimum dilution to form martensite on the Schaeffler diagram; ER308 had the lowest minimum dilution to form martensite and the most severe cracking, while ERNiCr-3 had the highest minimum dilution to form martensite and the least severe cracking.

Microstructures near the fusion boundary cannot be predicted on the Schaeffler diagram using the bulk weld metal dilution because of locally increased dilution near the fusion boundary. The tie line between the base metal and filler metal will allow the microstructure of the fusion boundary transition region to be approximated.

Acknowledgments

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References


